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Local crystallography and stress voiding in Al–Si–Cu versus copper interconnects

R. R. Keller,[a] C. E. Kalnas, and J. M. Phelps
National Institute of Standards and Technology, Materials Reliability Division, 325 Broadway, Boulder, Colorado 80303

(Received 11 January 1999; accepted for publication 12 April 1999)

We compare the local crystallographic orientations associated with stress voids in Al–1Si–0.5Cu (wt %) with those in pure copper interconnects. Orientations were sorted by whether grains were immediately adjacent to voids. Grains adjacent to voids in Al–Si–Cu showed a ⟨111⟩ fiber texture that was slightly stronger than those in intact regions. This is in contrast to copper, which showed weaker local ⟨111⟩ texture around voids. We postulate the difference to be due to the relative effectiveness of the diffusion paths available in the lines. For Al–Si–Cu, the presence of defects associated with precipitates may allow more rapid diffusion than grain boundaries. Voiding in copper, which is free from such defects, depends more on grain boundary structure. [S0021-8979(99)03814-1]

We present in this letter a comparison of local crystallography measurements associated with stress voiding in Al–Si–Cu versus those obtained in an earlier study[1,2] on pure copper interconnects. Although local variations in texture and grain boundary structure appear to have important effects on interconnect performance,[2,3] considerable work remains to be done to fully understand the effects of local texture on reliability. Kordic *et al.*[4] showed using electron backscatter diffraction (EBSD) that the degree of voiding in narrow lines could be correlated to the global texture strength. Lines of weaker average texture showed more voiding. While it has been often viewed that stronger global texture within interconnects leads to superior resistance to stress voiding and electromigration, this is not always the dominating factor in controlling voiding resistance. For example, Rodbell *et al.*[3] showed that stress voiding in aged Al–2Si–0.5Cu lines of stronger overall texture was more prevalent than that in lines of weaker overall texture. However, voids in the strongly textured lines formed near grains that were oriented significantly far from the average ⟨111⟩ texture. Similarly, Nucci *et al.*[1,2] showed that voiding in copper occurred near grains with orientations more widely distributed about the overall ⟨111⟩ texture than those in regions that remained intact. Conclusions from these investigations suggested that the presence of a relatively small population of grain boundaries of high diffusivity within the film plane can be detrimental to voiding resistance since metal atoms can then be easily transported away from a growing void at the metal/passivation interface.

Stress voiding in alloyed lines may also show similar dependence on the local variations in grain boundary structure, but only if more efficient diffusion paths are absent. If the material contains microstructural features that do provide such paths, the behavior may be different. For instance, Kordic *et al.*[5] showed that a network of dislocations surrounding silicon precipitates in Al–Si–Cu alloys can act to enhance material transport away from a void, leading to rapid growth. Our work addresses whether such alloys exhibit voiding that is also dependent on grain boundary structure, as has been observed for pure copper. Since the observations from copper have been reported earlier,[1,2] we refer to those results only as needed in order to provide context for the aluminum alloy observations.

Ti/Al–1Si–0.5Cu/TiN (wt %) lines were fabricated on thermally oxidized silicon (001) substrates at 250 °C. The lines were 0.4 μm thick and 2.0 μm wide. They were passivated by 2 μm of chemical-vapor-deposited SiO_2 deposited at 400 °C, and cooled to room temperature. Pure copper lines were fabricated as described in Ref. 6. They were deposited by electron beam evaporation into geometries defined by lift-off methods. The copper lines were 0.5 μm thick and 2.0 μm wide. Focused ion beam (FIB) imaging of lines in both metals revealed that the grains were columnar and that two to three grains were present across the linewidths. Prior to scanning electron microscope (SEM) observations, samples were depassivated by reactive ion etching to allow EBSD observations to be made.

An optical micrograph showing the distribution of stress voids in the aluminum alloy lines is shown in Fig. 1. From such micrographs, the number of voids in several areas was determined both before and after depassivation. Table I summarizes the void statistics for the lines. The void densities before and after depassivation of the alloyed lines are comparable, so it is reasonable to conclude that this process did not induce additional voids, nor did it begin dissolving the

[a]Electronic mail: bob.keller@nist.gov

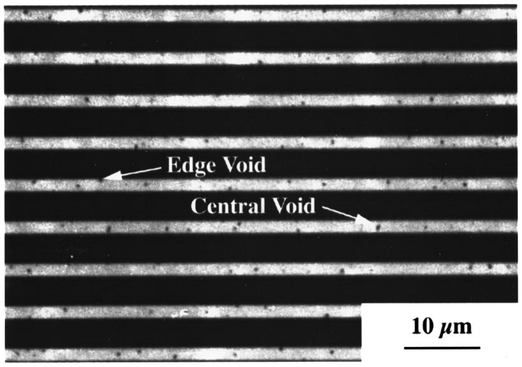

FIG. 1. Optical micrograph taken from depassivated Al–Si–Cu sample, showing distribution of stress voids. Edge and central void examples are indicated.

TABLE I. Stress void statistics in Al–Si–Cu and Cu lines.

Material	Total line length (μm)	Number of voids			Total void density (voids/μm^2)
		Total	Edge	Central	
Passivated Al–Si–Cu	2177	244	74	170	0.056
Depassivated Al–Si–Cu	1693	169	50	119	0.050
Depassivated Cu (1)[a]	700	91	11	80	0.065
Depassivated Cu (2)[a]	700	23	8	15	0.016

[a]Reference 6.

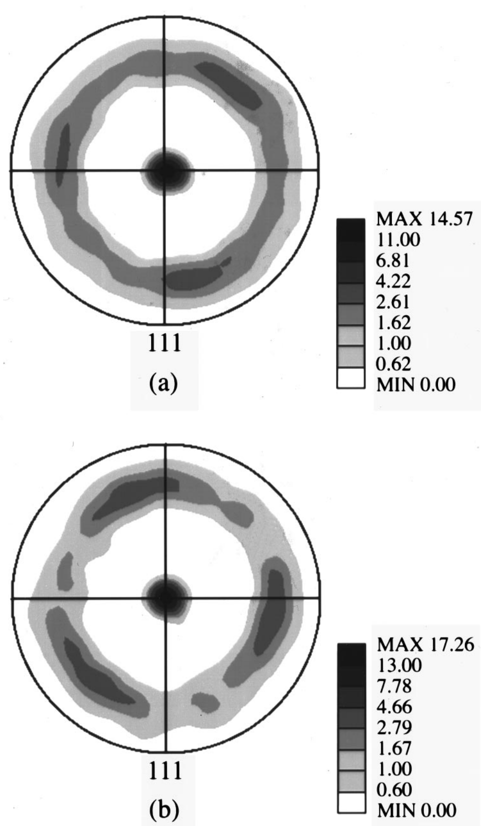

FIG. 2. (111) pole figures from (a) Al–Si–Cu, intact regions, (b) Al–Si–Cu, grains adjacent to voids. Scales are given in terms of times random ⟨111⟩ orientation distributions.

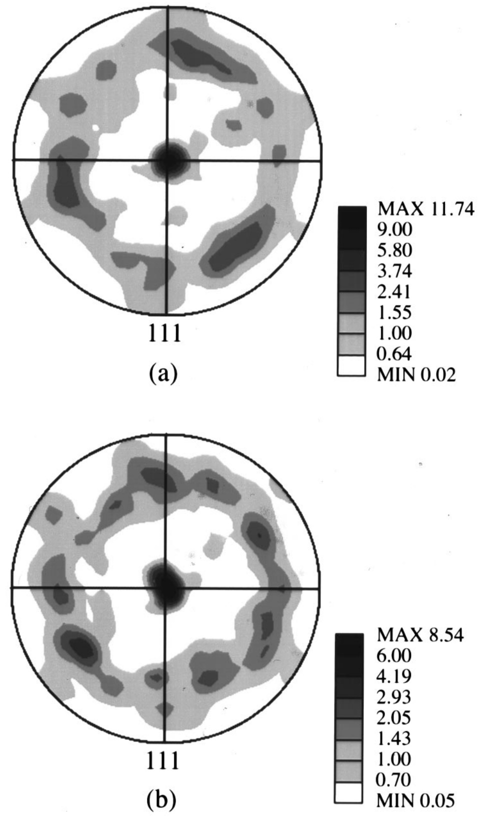

FIG. 3. (111) pole figures from (a) pure copper, intact regions, and (b) pure copper, grains adjacent to voids. Scales are given in terms of times random ⟨111⟩ orientation distributions.

metal. For comparison, the void densities for depassivated pure copper are also given;[6] in that work, Cu (1) had a slightly weaker global texture than Cu (2) and exhibited more voiding. The void densities for the aluminum alloy and Cu (1) are similar, and that for Cu (2) is lower by a factor of approximately 3–4. FIB results showed that voids between the line edges, or central voids, were located at triple junctions. The grains surrounding voids appeared to be no larger or smaller than those in intact regions. Voids at line edges were observed where grain boundaries intersected the line edge.

After depassivation, crystal orientations were measured using EBSD. The SEM was operated with an accelerating voltage in the range 15–30 kV, probe currents in the range

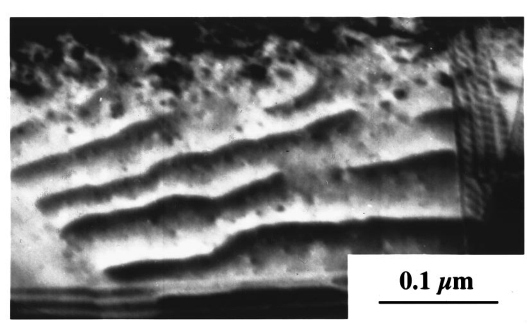

FIG. 4. Bright field transmission electron micrograph showing clusters of parallel dislocations in Al–Si–Cu grains.

1.5–3.0 nA, and a specimen tilt of approximately 70°. The beam was placed in approximately 6 to 7 positions around each of 29 voids in the aluminum alloy, as near to the void edges as possible; we estimate that on average the beam was placed within 0.2–0.3 μm of the void edges. Away from the voids, the beam was randomly placed with adjacent positions spaced no nearer than several micrometers. A total of 200 patterns were collected adjacent to voids and 205 patterns were collected far from voids.

Orientations associated with measurements adjacent to voids and those within intact regions of the aluminum alloy were summarized into (111) pole figures, which are shown in Fig. 2. The ⟨111⟩ fiber texture strength for grains adjacent to voids is somewhat (approximately 18%) higher than that for intact regions. For comparison, the ⟨111⟩ fiber texture strengths for grains adjacent to voids in copper were about 27% lower than that for intact regions for a similar number of measurements, as shown by the times random orientation distribution values in Fig. 3.

The observation of a higher ⟨111⟩ texture strength for the voided as compared to the intact regions for the aluminum alloy was at first somewhat surprising. Similar experiments performed on pure copper[1,2] lines indicated the opposite effect, namely, that grains near voids were not as likely to be ⟨111⟩ oriented as grains within intact regions. The discussion for such behavior in copper centered on the necessarily different grain boundaries associated with different texture strengths. The columnar structure of the films allowed approximate inferences to be made regarding grain boundary structure. Strong fiber texture in columnar grains restricts the possible grain boundaries to those that exhibit strong tilt character with misorientation axes normal to the film plane. In-plane diffusion through such boundaries is very limited and void growth should be quite slow. Weaker fiber texture allows for boundaries of twist character and those of tilt character with misorientation axes in the film plane. Such boundaries should allow for more efficient in-plane diffusion. The presence of long, rapid diffusion paths might be expected to result in larger voids after a given time than the absence of such paths.

It is apparent that off-⟨111⟩ grains in the aluminum alloy have a much smaller effect on voiding than such grains in pure copper. The most reasonable explanation for our observation is that there are microstructural features in the lines that provide diffusion paths more efficient than those provided by grain boundaries alone. Possible sources of such diffusion paths are silicon particles and Al–Cu precipitates surrounded by dislocation networks, such as those observed in Ref. 5. If a large number of such defects are present in the lines, and those defects can transport significant numbers of vacancies or atoms to or from a nucleated void, then the effects of local grain boundary structures could become diminished. This may explain why there is such selectivity in void locations in the earlier studies[1,2] as compared to this work. Transmission electron microscopy observations are in progress to confirm this hypothesis. Preliminary observations show that clusters of nearly parallel dislocations are sometimes present in grains viewed in transverse cross section as shown in Fig. 4; whether these are associated with silicon particles or other precipitates remains to be determined. It may be argued that the dominant diffusion path is simply through the passivation/metal interface. However, the fact that we saw voids only at triple junctions and grain boundaries suggests that the boundaries do indeed still play a role. That role may be, in the presence of high-diffusivity defect structures, to provide a distribution of nucleation sites that become either supplied with vacancies or depleted of atoms during void growth. The main conclusions we can draw are that stress void growth depends on the dominant local microstructural features controlling in-plane diffusion, and that those features are not necessarily just the grain boundaries.

The authors acknowledge the NIST Office of Microelectronics Programs and the National Research Council Post-Doctoral Research Associateship Program for support. They also thank J. E. Sanchez, Jr. for supplying specimens, and J. Bonevich of NIST and L. Petrosino of Micrion Corp. for focused ion beam work.

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